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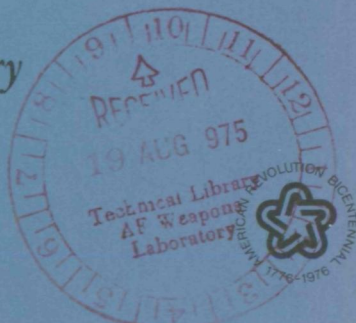
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THERMAL EXPANSION IN THE
NICKEL-CHROMIUM-ALUMINUM AND
COBALT-CHROMIUM-ALUMINUM SYSTEMS TO 1200° C

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16. Abstract <p>Thermal expansion data were obtained on 12 Ni-Cr-Al and 9 Co-Cr-Al alloys by high-temperature X-ray diffraction. The data were computer fit to an empirical thermal expansion equation developed in this study:</p> $LP_T = LP_{250C}(1 + R)[1 + (T/273)]^{1.5}$ <p>where LP_T is the lattice constant at any temperature T in °C, LP_{250C} is the lattice constant at 25° C, and R is an expansion constant. The fit was excellent to good. The expansion constants depended on phase but not on composition. For phases in the Ni-Cr-Al system, the expansion constants were as follows: 19.2×10^{-4} for γ/γ', 19.9×10^{-4} for β, and 13.4×10^{-4} for α-Cr. For phases in the Co-Cr-Al system, the expansion constants were as follows: 20.9×10^{-4} for α-Co and 17.8×10^{-4} for β. Only α-Cr had an expansion constant low enough to minimize oxide spalling or coating cracking induced by thermal expansion mismatch.</p>					
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THERMAL EXPANSION IN THE NICKEL-CHROMIUM-ALUMINUM AND COBALT-CHROMIUM-ALUMINUM SYSTEMS TO 1200° C

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SUMMARY

The effect of temperature on the lattice parameters of phases in 12 nickel-chromium-aluminum (Ni-Cr-Al) alloys and 9 cobalt-chromium-aluminum (Co-Cr-Al) alloys was determined by high-temperature X-ray diffraction (HTXRD). The temperature range was 25° to 1200° C. The data for each phase of each alloy were computer fit to an empirical thermal expansion equation developed in this study:

$$LP_T = LP_{25^{\circ}C} (1 + R) [1 + (T/273)]^{1.5}$$

where LP_T is the lattice constant at any temperature T in °C, $LP_{25^{\circ}C}$ is the lattice constant at 25° C, and R is an expansion constant. The fit was excellent in nearly all cases. An expansion constant was derived for each phase. Comparing expansion constants revealed that, for a given phase, R was independent of alloy composition. For phases in the Ni-Cr-Al system, the expansion constants were as follows: 19.2×10^{-4} for γ/γ' , 19.9×10^{-4} for β , and 13.4×10^{-4} for α -Cr. For phases in the Co-Cr-Al system, the expansion constants were as follows: 20.9×10^{-4} for α -Co and 17.8×10^{-4} for β . Of all the phases, only α -Cr in the Ni-Cr-Al system had an R sufficiently low to reduce to an unimportant level the stress induced by thermal expansion mismatch between oxide and substrate or coating and substrate.

INTRODUCTION

The nickel-chromium-aluminum (Ni-Cr-Al) and cobalt-chromium-aluminum (Co-Cr-Al) systems, commonly called M-Cr-Al systems, are becoming increasingly important in high-temperature applications where combined oxidation and hot corrosion resistance are required. Such applications include coatings for blades and vanes in ad-

vanced gas turbine engines (ref. 1) or matrices for dispersion-strengthened alloys in similar turbine engine applications (ref. 2). M-Cr-Al systems are highly oxidation resistant because of their initial rapid formation of a thin protective oxide layer (mostly aluminum sesquioxide (Al_2O_3)) upon high-temperature exposure. In intermittent service, however, there is a tendency for this protective alumina-rich scale to spall off during each cooling cycle. Such spalling requires a renewed formation of the oxide upon heating. After many such heating and cooling cycles, alloy surfaces become sufficiently depleted in aluminum (unless resupplied) so that other less protective oxides form (ref. 3). The oxidation process is then accelerated until ultimately the material is so severely attacked that it is subject to surface-induced failure in service.

The most commonly accepted cause for the spalling of oxides during thermal cycling can be described as follows:

- (1) The oxide scale forms at temperature and is coherent with the substrate.
- (2) As the material cools at the end of a cycle, the oxide and the metal contract differently.
- (3) This results in a stress in the substrate and a stress in the oxide.
- (4) When the stress in the oxide exceeds its strength, it fails and the oxide cracks and/or spalls off.

If this mechanism is the primary cause of spalling, reducing the difference in the coefficients of thermal expansion (CTE) between the substrate and the oxide would lessen the stress in the oxide and hence decrease the tendency for the oxide to spall. A similar argument has been used to explain M-Cr-Al coating failure by cracking. Here the important difference in CTE is between the M-Cr-Al-based coating and the substrate.

The work described in this report is part of a larger program designed to identify the optimum M-Cr-Al composition ranges based on a number of evaluation criteria.

The program includes work in the following areas:

- (1) Cyclic oxidation
- (2) Cyclic hot corrosion
- (3) Coefficient of thermal expansion (CTE)
- (4) Ductility
- (5) Diffusion

In this report the CTE's in the Ni-Cr-Al and Co-Cr-Al systems are surveyed for those compositions which show the greatest potentials for combined oxidation and hot corrosion resistance. To this end, the thermal expansion of individual phases for a number of single and multiphased materials (12 Ni-base and 9 Co-base) were determined by high-temperature X-ray diffraction (HTXRD). The data were then fit by computer to a thermal expansion equation which allowed the CTE for each phase of each alloy to be characterized by a single constant. These constants were then assessed to determine whether or not optimum alloys could be chosen to minimize the CTE effect in spalling and/or coating cracking.

MATERIALS AND SAMPLE PREPARATION

The distributions of compositions selected for casting are shown in figures 1 and 2. The original intent was to cast nine different compositions in each system ranging from 6- to 30-at. % Al and 10- to 22-at. % Cr. (Several Ni-base alloys had to be recast to get closer to these nominal compositions.) The distribution was chosen to cover the compositions expected to have good oxidation and corrosion resistance. A "star" array was selected to give the most information on the effects of trends in composition with the least number of data points (ref. 4). All alloys were melted in zirconia crucibles and cast under argon into zirconia molds. Each casting consisted of a "tree" of 10 coupons whose dimensions were 2.5 cm by 5.1 cm by 0.25 cm. Each coupon had a small triangular riser attached which was removed after casting and used for chemical analysis by atomic absorption spectroscopy. The results of these analyses for Cr and Al are shown in table I.

A coupon from each casting was cut into several 1.3-cm by 0.95-cm by 0.25-cm pieces and glass-bead blasted. These coupons were used to make the expansion measurements.

EXPERIMENTAL PROCEDURES

The high-temperature X-ray diffractometer (HTXRD) has been described fully in several publications (e. g., ref. 5). In summary, each sample was mounted in the diffractometer, briefly heated in helium to 1200° C, and slowly cooled to room temperature before measuring in order to relieve stress from the glass-bead blasting. A complete diffractometer scan was made at room temperature in order to identify the phases present and to obtain room-temperature lattice constants. The samples were reheated to 50° C and then heated in 50° C increments to 1200° C. At each temperature a lattice constant was determined for each phase by using the highest 2 θ diffraction lines possible. The samples were then cooled in 50° C increments. Lattice constants were obtained at each temperature to a precision of ± 0.0001 nm (± 0.001 Å). Upon cooling to room temperature, another complete scan was made to check for possible changes in the phases or for excessive oxidation.

When all lattice constant data were collected, the lattice constant values for each phase of each alloy were computer fit to an equation developed for this study:

$$LP_T = LP_{250C}(1 + R)[1 + (T/273)]^{1.5} \quad (1)$$

where LP_T is the lattice constant at any temperature T in °C, LP_{250C} is the lattice

constant at 25⁰ C, and R is an expansion constant. Also calculated was the mean coefficient of thermal expansion (CTE) over the temperature range 25⁰ to 1200⁰ C. This procedure is detailed in appendix A.

RESULTS

Exposing the test alloys to 1200⁰ C in static helium resulted in only minor amounts of oxides being formed from residual oxygen in the gas and possible leaks in the system. The oxidation that did take place had little effect upon the lattice parameter data, which showed hysteresis for only one alloy, Ni - 19-at. % Cr - 24-at. % Al. In all cases the alloys gave the same lattice constants before and after the CTE run. The phases that were found and for which CTE's were determined in the Ni-Cr-Al system were γ (nickel solid solution), γ' (Ni₃Al type), β (NiAl type), and α -Cr (chromium solid solution). This was expected from the phase diagram (ref. 6). When γ' is present in an alloy, the presence of γ cannot be determined by X-ray diffraction. The reason is that the diffraction pattern of γ' contains all the lines of γ plus a few extra lines. Both phases have the same lattice constant and apparently the same CTE because no splitting of the diffraction lines was seen even at 1200⁰ C. Only the α -Co (Co solid solution) and β (CoAl type) phases were found in the Co-Cr-Al system. This may, in part, account for the greater precision found in the Co-Cr-Al data.

Figures 3 to 6 are representative of the data obtained for the solid solution (γ , γ' , and α -Cr) and NiAl type (β) phases of the Ni- and Co-Cr-Al alloys. Even when a phase was present over a limited temperature range (fig. 7), excellent fits, which would be consistent with more complete data sets, were obtained and the scatter was small. In the very few cases where the scatter was larger (as in fig. 8), a reasonable fit and expansion constant were found.

Tables II and III contain values for LP_{25^0C} , R, and mean CTE from 25⁰ to 1200⁰ C for all phases in each alloy. For the Ni-Cr-Al system, multiple linear regression combined with analysis of variance (ref. 7) showed that, with a rejection level of 0.05, expansion constants for γ/γ' (average 19.2×10^{-4}) and β (average 19.9×10^{-4}) were not composition dependent. The same analysis showed the expansion constants for γ/γ' and β to be the same, while those for α -Cr (average 13.4×10^{-4}) differed from those for either γ/γ' or β . The analysis is detailed in appendix B. In the Co-Cr-Al system there is a significant difference between the expansion constants of α -Co and β , 20.9×10^{-4} and 17.8×10^{-4} , respectively.

DISCUSSION

Thermal expansion data and their implications are discussed both generally and as they apply to oxidation spallation and coating cracking. In general, the various methods of expressing thermal expansion data are compared. Also the expansion coefficients obtained in this investigation are compared with previously published data.

General

There is no generally accepted method of presenting thermal expansion data. They are often expressed as the polynomial

$$LP_T = LP_0(1 + AT + BT^2 + CT^3 \dots) \quad (2)$$

although the number of terms varies from two to more than five (ref. 8). This approach often leads to excellent approximations of the data, but it has the disadvantage of making it difficult to compare one material to another since several constants are involved. In addition, extrapolation from limited data sets is very unreliable. Another widely used method is to reduce the data to a mean CTE. This has the advantage of yielding a single constant for ease of comparison, but the value is valid only for the quoted temperature range and gives no clue as to the shape of the thermal expansion curve.

The method used in this investigation has the advantages of both these techniques but none of their drawbacks. It allows a material's expansivity to be expressed by one constant, R . Also it describes the shape of the original lattice constant - temperature curve even with limited data sets.

That this technique gives data similar to those obtained by conventional dilatometric techniques can be shown by comparing these data with current work by Felten, Friedrich, and Strangman of Pratt & Whitney Aircraft under contract NAS3-18920. Their mean CTE at 1200° C for Ni-18Cr-5Al and Ni-18Cr-9Al are (extrapolated from 1100° C) 20.7 and 18.2, respectively. These values are in reasonable agreement with the data obtained by HTXRD.

One source of concern in using HTXRD is that the values obtained for the individual phases might be influenced by interphase constraint. That is, phases with differing expansion coefficients might prevent each other from expanding in the same amount as they would as single phases. However, this effect was not found. Where the HTXRD data could be compared (at 20° C) with literature values, as for α -Cr, α -Co, and Ni (ref. 9), little difference was noted. In addition, this effect should be a function of the relative amount of the phases. A glance at tables II and III shows no evidence of such interactions.

Assessment of Specific Results

Spalling and CTE mismatch. - There are several equations that can be used for calculating the stress induced in an oxide scale from thermal expansion mismatch (refs. 10 to 12). However, all give similar results. The equation quoted by Douglass (ref. 11) is

$$\sigma_{ox} = \frac{E_{ox} \Delta T (CTE_{ox} - CTE_m)}{1 + 2 \frac{E_{ox}}{E_m} \left(\frac{t_{ox}}{t_m} \right)} \quad (3)$$

where

σ_{ox}	stress in oxide, N/m^2
E_{ox}	elastic modulus of oxide, N/m^2
ΔT	difference between oxidizing temperature and temperature to which sample is cooled, $^{\circ}C$
CTE_{ox}	CTE of oxide over ΔT , $^{\circ}C^{-1}$
CTE_m	CTE of metal over ΔT , $^{\circ}C^{-1}$
E_m	elastic modulus of metal, N/m^2
t_{ox}	thickness of oxide, cm
t_m	thickness of metal, cm

The primary assumptions are that only CTE controls spalling and that neither thermal shock nor growth stresses are important. For oxidation of Ni-Cr-Al, $t_{ox} \ll t_m$ and $E_{ox}/E_m < 2$. Therefore,

$$1 + 2 \frac{E_{ox}}{E_m} \left(\frac{t_{ox}}{t_m} \right) \approx 1$$

If the oxidation temperature is $1200^{\circ}C$ and the material is cooled to room temperature ($25^{\circ}C$), ΔT equals $1175^{\circ}C$. Since the most protective oxide is Al_2O_3 , which is known to form on M-Cr-Al systems, its elastic modulus in bulk form will be used ($37 \times 10^{10} N/m^2$, ref. 13). Equation (3) then becomes

$$\sigma_{ox} = -43 \times 10^{13} (CTE_{ox} - CTE_m) \quad N/m^2 \quad (4)$$

As CTE_m is greater than CTE_{ox} , the oxide is in compression. Then putting in the compressive strength of a solid body of Al_2O_3 ($31 \times 10^8 \text{ N/m}^2$, ref. 13) allows the calculation of the maximum allowable change in thermal expansion coefficient ΔCTE_{max} to avoid failure of the oxide:

$$\Delta CTE_{max} = -7.1 \times 10^{-6} \text{ } ^\circ\text{C}^{-1}$$

The value for CTE_{ox} is $8.1 \times 10^{-6} \text{ } ^\circ\text{C}^{-1}$ (ref. 14). Therefore, the maximum CTE_m is $15 \times 10^{-6} \text{ } ^\circ\text{C}^{-1}$. In tables II and III only the α -Cr phase in the M-Cr-Al systems is below this value. As far as CTE is concerned, the way to reduce spallation in the Ni-Cr-Al system is to produce alloys with a high volume fraction (at least 67 percent) of α -Cr. There seems to be little possibility of doing this in the Co-Cr-Al system.

Coating, cracking, and CTE mismatch. - When the effect of CTE mismatch on a coating-substrate system is being considered, the calculations are a little more nebulous. The ratio of thicknesses is not insignificant and the elastic moduli and strengths of M-Cr-Al systems are not known. For a coating-substrate system, the subscripts ox and m become coat (coating) and subs (substrate), respectively. Assuming a modulus of $21 \times 10^{10} \text{ N/m}^2$ and a t_{coat}/t_{subs} of 0.1 yields a stress in the coating of

$$\sigma_{coat} = 18 \times 10^{10} (\Delta T)(\Delta CTE) \text{ N/m}^2 \quad (5)$$

Coating deposition temperatures range from 900° to 1100° C ; an average ΔT may be taken to be 1000° C . In this case, equation (5) reduces to

$$\sigma_{coat} = 18 \times 10^7 \Delta CTE \text{ N/m}^2$$

where ΔCTE is in units of $10^{-6} \text{ } ^\circ\text{C}^{-1}$.

For example, at 1000° C , the CTE of an experimental blade material (a directionally solidified eutectic), $\gamma/\gamma' - \delta$, is $14 \times 10^{-6} \text{ } ^\circ\text{C}^{-1}$ (Pratt & Whitney Aircraft, NAS3-18920). This CTE is below those of all phases of the M-Cr-Al systems except α -Cr of Ni-Cr-Al system and is due to the large volume fraction of Ni_3Nb in this alloy. Therefore, M-Cr-Al coatings would be in tension, except for those with very high amounts of α -Cr. Because its CTE is slightly lower than that of $\gamma/\gamma' - \delta$, coatings rich in α -Cr would be in slight compression. Without better knowledge of the moduli and yield strengths of M-Cr-Al systems, nothing further can be decided about the CTE effect on coatings. In summary, the α -Cr phase apparently could minimize thermal stresses in a coating on a low-CTE alloy and could reduce spalling of aluminum oxide.

SUMMARY OF RESULTS

The results of using high-temperature X-ray diffraction (HTXRD) to determine the coefficients of thermal expansion (CTE) of Ni-Cr-Al and Co-Cr-Al alloys from room temperature to 1200^o C may be summarized as follows:

1. Expansion data for these systems can be well described by an equation developed in this study with only one constant, R, for a lattice parameter at any temperature:

$$LP_T = LP_{25^oC} (1 + R) [1 + (T/273)]^{1.5}$$

2. The expansion constants R for γ/γ' and β in the Ni-Cr-Al system were nearly the same, 19.2×10^{-4} and 19.9×10^{-4} , respectively.

3. The expansion constant R for α -Cr was 13.4×10^{-4} .

4. There was a slight but significant difference between the expansion constants for α -Co and β in the Co-Cr-Al system, 20.9×10^{-4} and 17.8×10^{-4} , respectively.

CONCLUSIONS

From the thermal expansion data obtained on alloys in the Ni- and Co-Cr-Al systems, the following conclusions may be drawn:

1. The α -Cr phase in the Ni-Cr-Al alloys is most desirable for minimizing thermal stresses induced by thermal expansion coefficient (CTE) mismatch either between an oxide and an alloy or between $\gamma/\gamma' - \delta$ and an M-Cr-Al coating.

2. In the Co-Cr-Al system, little can be done to minimize CTE mismatch because the reduction in going from α -Co to β would be slight.

3. The equation $LP_T = LP_{25^oC} (1 + R) [1 + (T/273)]^{1.5}$, where LP is the lattice parameter at any temperature T, LP_{25^oC} is the lattice parameter at 25^o C, and R is an expansion constant, is most useful in describing thermal expansion and comparing thermal expansion data. It fully describes the shape of the curve and allows thermal expansion data to be compared by using a single coefficient for each material.

Lewis Research Center,

National Aeronautics and Space Administration,

Cleveland, Ohio, May 9, 1975,

505-01.

APPENDIX A

DEVELOPMENT OF MATHEMATICAL MODEL

In scanning the data, one of many reasonable candidates for an appropriate mathematical model appeared to be

$$y = A(B)^x \quad (A1)$$

where $y = LP_T$ and $x = T$. In this case the parameters A and B could be estimated by linearizing the model in the form

$$\ln y = \ln A + x \ln B$$

and performing a linear regression.

In observing the results, two problems became apparent. The first difficulty arose from the fact that the dependent-variable differences were several magnitudes smaller than those of the independent variable. As a result, the parameter $\ln B$ hovered near zero but remained positive. This led to the observation that a compound growth equation

$$y = A(1 + r)^x \quad (A2)$$

was, in fact, a model that was more realistic and that could better cope with the problem of scaling since it could be treated as

$$y = A(1 + qr)^{x/q}$$

where r is the rate of growth and q could be chosen a priori. Since x was given in $^{\circ}\text{C}$, a first choice for q was 273. This choice proved to be successful. It could be used both as a scale factor and as a constant to convert from Celsius to Kelvin. Thus, the independent variable became

$$\frac{x}{q} = \frac{T + 273}{273}$$

The second problem hinged on the need to provide an improved fit. The fit was improved by raising the independent variable to some power. Since many physical relationships depend on $T^{3/2}$, an exponent of 1.5 seemed to be the most likely candidate.

These modifications resulted in the model

$$y = A(1 + 273)^{[1+(T/273)]^{1.5}}$$

which could be linearized to

$$\ln y = \ln A + x^{1.5} \ln B$$

or in terms of this study

$$\ln LP_T = \ln LP_{25^\circ C} + \left(1 + \frac{T}{273}\right)^{1.5} \ln(1 + R)$$

or

$$LP_T = LP_{25^\circ C} (1 + R)^{[1+(T/273)]^{1.5}}$$

The least-squares method could then be used to determine the parameters $\ln A$ and $\ln B$. In addition, error estimates of the parameters as well as an overall standard deviation were calculated for each set of data.

APPENDIX B

MULTIPLE LINEAR REGRESSION WITH ANALYSIS OF VARIANCE

A multiple linear regression analysis was performed on the expansion constants for Ni-Cr-Al listed in table II according to the model equation

$$R = b_0 + b_1x_1 + b_2x_2 \pm \text{S. E. E.} \quad (\text{B1})$$

by using the technique described in reference 7 with dummy variables. In the present investigation,

Phase	x_1	x_2
γ/γ'	0	0
β	1	0
α	0	1

Because there were two duplicate runs available, these were used for the mean-square error terms. The final equation for an α rejection level of 0.05 was

$$R = 19.46 - 6.06 x_2 \pm 1.43$$

The hypothesis that all the γ/γ' expansion constants are the same was tested by the lack-of-fit term in the following analysis of variance (ANOVA) table:

Source	Sum of squares	Degrees of freedom	Mean square
Lack of fit	44.4045	19	2.2202
replication	.5650	2	.2825
Total re-sidual	44.9695	--	2.0441

The standard estimate of error is $(2.0441)^{1/2}$ and the variance ratio test is

$$F = \frac{\text{Mean-square lack of fit}}{\text{Mean-square replication}} = 7.859$$

At the 0.05 significance level, the value of F for 19 and 2 degrees of freedom should exceed 19.4. Therefore, the lack-of-fit term, which is a measure of differences in R , is not significant (i.e., all R 's are equal).

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TABLE I. - CHEMICAL ANALYSIS OF AS-CAST MATERIALS

Alloy	Base	Chromium content,		Aluminum content,	
		at. %	wt %	at. %	wt %
1	Nickel	15.98	15.96	17.54	9.09
2A	↓	11.50	12.00	25.58	13.85
2B		12.44	12.77	22.72	12.10
3		13.19	12.70	12.07	6.03
4		18.41	17.74	11.06	5.53
5A		14.35	14.85	23.65	12.70
5B		16.89	18.18	29.19	16.30
6		19.15	20.01	24.16	13.10
7		15.81	14.73	5.77	2.79
8A	↓	18.87	20.98	26.99	14.90
8B		20.84	20.81	16.52	8.56

Alloy	Base	Chromium content,		Aluminum content,	
		at. %	wt %	at. %	wt %
9	Nickel	9.73	9.62	17.18	8.81
10	Cobalt ↓	15.86	15.86	18.30	9.50
11		11.55	11.68	20.97	11.00
12		12.85	12.33	12.05	6.00
13		18.78	18.11	11.59	5.80
14		18.93	19.61	23.25	12.50
15		10.46	10.38	18.25	9.04
16		13.07	13.54	24.55	13.20
17		15.90	14.74	5.41	2.60
18		22.11	22.12	16.95	8.80

TABLE II. - THERMAL EXPANSION OF Ni-Cr-Al ALLOYS

Alloy, at. %	Phase											
	γ/γ'						β					
	Lattice parameter at 25°C, LP _{25°C'} nm	Expansion constant, R	Mean co- efficient of thermal expansion, CTE, °C ⁻¹	Standard deviation of lattice parameter, σ_{LP}	Lattice parameter at 25°C, LP _{25°C'} nm	Expansion constant, R	Mean co- efficient of thermal expansion, CTE, °C ⁻¹	Standard deviation of lattice parameter, σ_{LP}	Lattice parameter at 25°C, LP _{25°C'} nm	Expansion constant, R	Mean co- efficient of thermal expansion, CTE, °C ⁻¹	Standard deviation of lattice parameter, σ_{LP}
Ni-13Cr-12Al	0.3557	20.6×10 ⁻⁴	20×10 ⁻⁶	0.14×10 ⁻²	---	---	---	---	---	---	---	---
Ni-18Cr-11Al	.3563	20.5	20	.10	---	---	---	---	---	---	---	---
Ni-16Cr-6Al	.3545	18.6	18	.07	---	---	---	---	---	---	---	---
Ni-16Cr-6Al ^a	.3551	19.3	19	.09	---	---	---	---	---	---	---	---
Ni-10Cr-17Al	.3561	20.5	20	.23	---	---	---	---	---	---	---	---
Ni-16Cr-18Al	.3573	19.8	19	.12	0.2851	18.7×10 ⁻⁴	18×10 ⁻⁶	0.10×10 ⁻²	---	---	---	---
Ni-12Cr-26Al	.3574	19.8	19	.19	.2864	17.9	17	.17	---	---	---	---
Ni-12Cr-23Al	.3574	19.4	19	.08	.2852	18.4	18	.07	---	---	---	---
Ni-14Cr-24Al	.3574	17.6	17	.16	.2845	22.5	22	.03	---	---	---	---
Ni-14Cr-24Al ^a	.3578	16.8	17	.28	---	---	---	---	---	---	---	---
Ni-21Cr-17Al	.3572	20.5	20	.24	.2853	19.9	20	.06	---	---	---	---
Ni-19Cr-24Al	.3574	17.9	17	.11	.2862	19.1	19	.46	0.2880	13.9×10 ⁻⁴	13×10 ⁻⁶	0.20×10 ⁻²
Ni-19Cr-27Al	.3572	17.9	18	.14	.2860	20.6	20	.17	.2880	13.2	13	.06
Ni-17Cr-29Al	---	---	---	---	.2855	22.4	22	.20	.2879	13.1	13	.05
Average	---	19.4	---	---	---	19.9	---	---	---	13.4	---	---

^aReplicate specimens.

TABLE III. - THERMAL EXPANSION OF Co-Cr-Al ALLOYS

Alloy, at. %	Phase							
	α -Co				β			
	Lattice parameter at 25 ^o C, LP _{25^oC} , nm	Expansion constant, R	Mean co- efficient of thermal expansion, CTE, °C ⁻¹	Standard deviation of lattice parameter, σ_{LP}	Lattice parameter at 25 ^o C, LP _{25^oC} , nm	Expansion constant, R	Mean co- efficient of thermal expansion, CTE, °C ⁻¹	Standard deviation of lattice parameter, σ_{LP}
Co-13Cr-12Al	0.3565	20.0 $\times 10^{-4}$	20 $\times 10^{-6}$	0.07 $\times 10^{-2}$	-----	-----	-----	-----
Co-19Cr-12Al	.3569	20.7	20	.11	-----	-----	-----	-----
Co-16Cr-5Al	.3546	18.6	18	.07	-----	-----	-----	-----
Co-10Cr-18Al	.3567	20.1	20	.11	-----	-----	-----	-----
Co-16Cr-18Al	.3564	21.6	21	.20	0.2861	17.7 $\times 10^{-4}$	17 $\times 10^{-6}$	0.05 $\times 10^{-2}$
Co-12Cr-26Al	.3563	21.8	↓	.16	.2861	17.7	↓	.06
Co-19Cr-23Al	.3574	21.6		.20	.2864	17.8		.06
Co-13Cr-25Al	.3566	21.8		.20	.2861	18.0		.08
Co-22Cr-17Al	.3571	21.7		.15	.2865	17.9		.06
Average	-----	20.9	-----	-----	-----	17.8	-----	-----

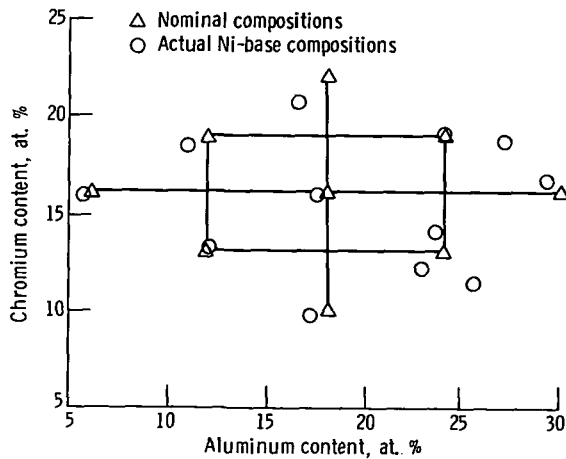


Figure 1. - Distribution of projected and actual casting compositions in Ni-Cr-Al system.

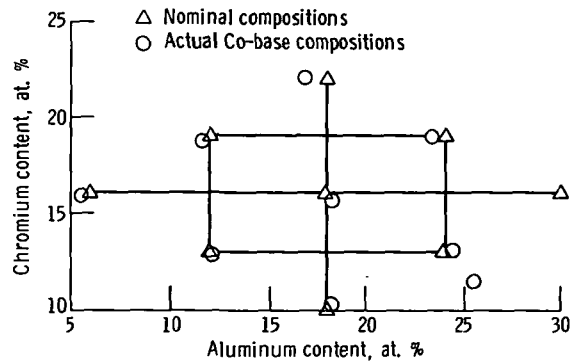


Figure 2. - Distribution of projected and actual casting compositions in Co-Cr-Al system.

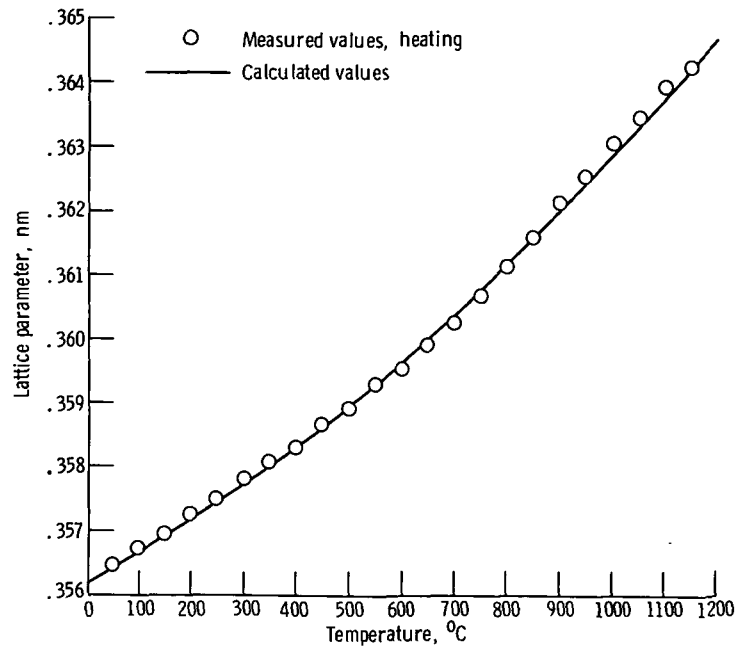


Figure 3. - Typical thermal expansion curve for γ/γ' in Ni-Cr-Al system (alloy Ni-18Cr-11Al). Standard deviation of lattice parameter, σ_{LP} , 0.0010.

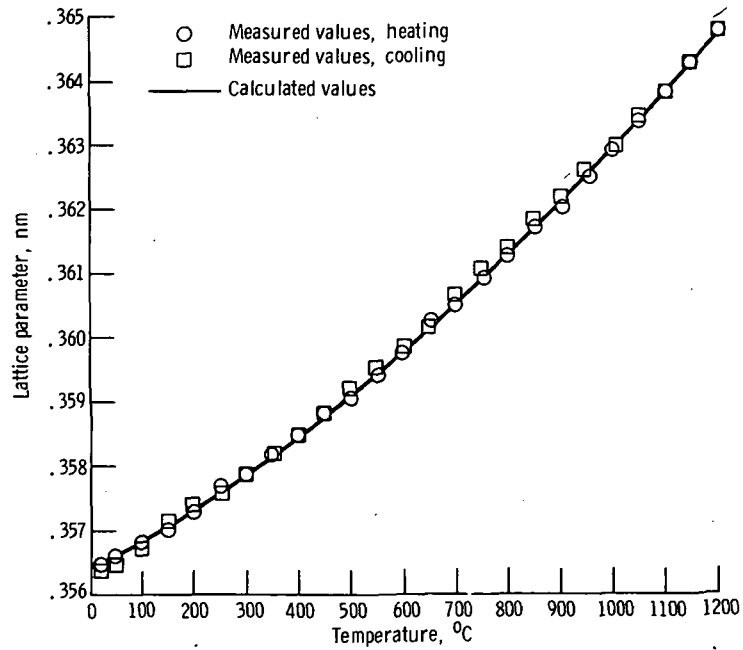


Figure 4. - Typical thermal expansion curve for α -Co in Co-Cr-Al system (alloy Co-13Cr-12Al). Standard deviation of lattice parameter, σ_{LP} , 0.0007.

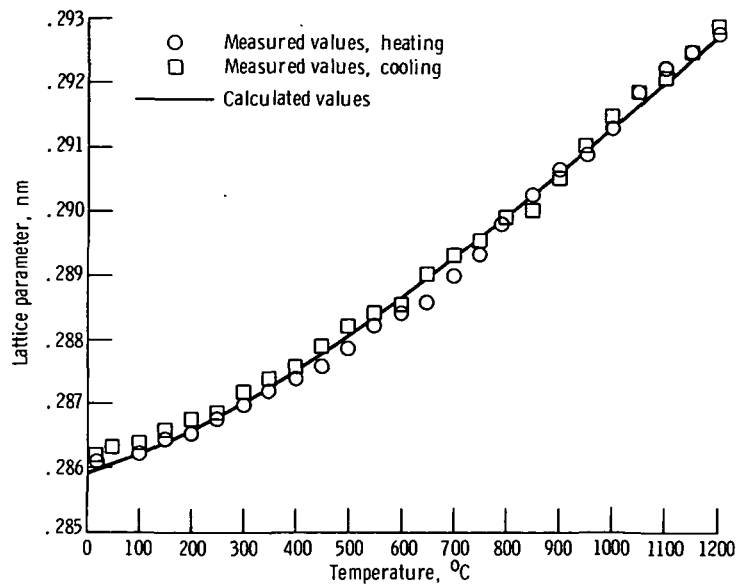


Figure 5. - Typical thermal expansion curve for β in Ni-Cr-Al system (alloy Ni-19Cr-27Al). Standard deviation of lattice parameter, σ_{LP} , 0.0011.

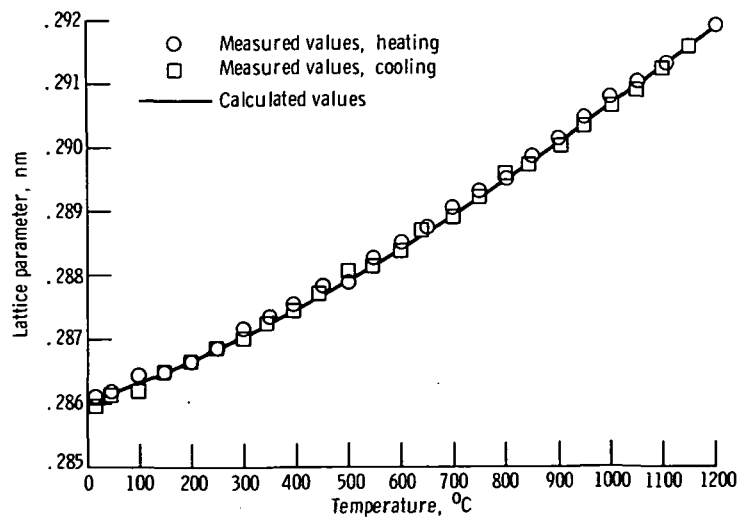


Figure 6. - Typical thermal expansion curve for β in Co-Cr-Al system (alloy Co-16Cr-18Al). Standard deviation of lattice parameter, σ_{LP} , 0.0005.

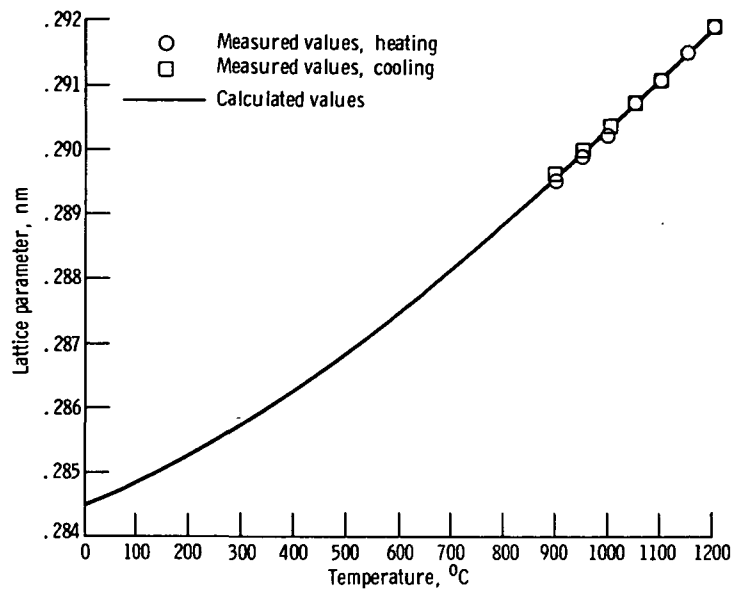


Figure 7. - Curve fitting with limited data - β in Ni-Cr-Al system (alloy Ni-17Cr-29Al). Standard deviation of lattice parameter, σ_{LP} , 0.0020.

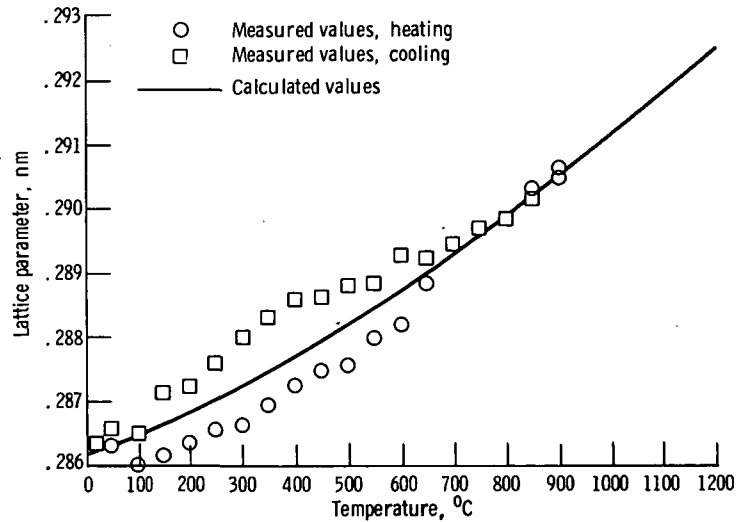


Figure 8. - Curve fitting to worst data set - β in Ni-Cr-Al system (alloy Ni-19Cr-24Al). Standard deviation of lattice parameter, σ_{LP} , 0.0046.



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